

UNCLASSIFIED

AD 407 394

DEFENSE DOCUMENTATION CENTER

FOR

SCIENTIFIC AND TECHNICAL INFORMATION

CAMERON STATION, ALEXANDRIA, VIRGINIA



UNCLASSIFIED

NOTICE: When government or other drawings, specifications or other data are used for any purpose other than in connection with a definitely related government procurement operation, the U. S. Government thereby incurs no responsibility, nor any obligation whatsoever; and the fact that the Government may have formulated, furnished, or in any way supplied the said drawings, specifications, or other data is not to be regarded by implication or otherwise as in any manner licensing the holder or any other person or corporation, or conveying any rights or permission to manufacture, use or sell any patented invention that may in any way be related thereto.

The Defense Metals Information Center was established at Battelle Memorial Institute at the request of the Office of the Director of Defense Research and Engineering to provide Government contractors and their suppliers technical assistance and information on titanium, beryllium, magnesium, refractory metals, high-strength alloys for high-temperature service, corrosion- and oxidation-resistant coatings, and thermal-protection systems. Its functions, under the direction of the Office of the Secretary of Defense, are as follows:

1. To collect, store, and disseminate technical information on the current status of research and development of the above materials.
2. To supplement established Service activities in providing technical advisory services to producers, melters, and fabricators of the above materials, and to designers and fabricators of military equipment containing these materials.
3. To assist the Government agencies and their contractors in developing technical data required for preparation of specifications for the above materials.
4. On assignment, to conduct surveys, or laboratory research investigations, mainly of a short-range nature, as required, to ascertain causes of troubles encountered by fabricators, or to fill minor gaps in established research programs.

Contract No. AF 33(616)-7747
Project No. 2(8-8975)

"The information in this report came from many sources, and the original language may have been extensively quoted. Quotations should credit the original authors and the originating agency. Where patent questions appear to be involved, the usual preliminary search is advised before making use of the material, and where copyrighted material is used, permission should be obtained for its further publication."

COPIES AVAILABLE FROM OTS \$

DMIC Report 182
January 31, 1963

**STRUCTURAL CONSIDERATIONS IN DEVELOPING
REFRACTORY METAL ALLOYS**

by

R. I. Jaffee and G. T. Hahn

to

**OFFICE OF THE DIRECTOR OF DEFENSE
RESEARCH AND ENGINEERING**

DEFENSE METALS INFORMATION CENTER
Battelle Memorial Institute
Columbus 1, Ohio

TABLE OF CONTENTS

	<u>Page</u>
SUMMARY	1
INTRODUCTION	3
LOW-TEMPERATURE STRENGTH	4
Resistance to Flow	4
Resistance to Fracture	10
The Rhenium Alloying Effect	14
INTERMEDIATE TEMPERATURES	21
HIGH-TEMPERATURE STRENGTH	21
CONCLUDING REMARKS	26
REFERENCES	27

STRUCTURAL CONSIDERATIONS IN DEVELOPING REFRACTORY METAL ALLOYS

SUMMARY

Progress made in applying advanced techniques and structural concepts to the problem of strengthening the Group VI-A metals, Cr, Mo, and W, is examined.

At low temperatures, $T/T_m < 0.2$, dislocations and grain boundaries play an important role. High strength values and the marked temperature and strain-rate dependence of the yield stress exhibited by the Group VI-A metals in this range are consistent with the idea that dislocation movement is rate controlling and thermally activated. Dislocation-mobility considerations also point to the existence of high stresses in advance of a moving crack, and, consequently, a low level of crack-propagation resistance. A second factor contributing to the brittleness of the Group VI-A metals is grain-boundary weakness caused by interstitial-atom impurities. Studies of rhenium additions to the Group VI-A metals suggest that the brittleness tendency can be inhibited by alloying. The origin of this effect has not been decided. There is evidence that alloying with rhenium influences dissolved interstitial atoms, changes the morphology of dislocations and grain-boundary oxides, promotes twinning, and affects the density of states at the Fermi level. These concepts, it is noted, provide a useful rationale of mechanical strength but have thus far yielded little in the way of practical alloys.

At intermediate temperatures, $0.2 < T/T_m < 0.4$, dislocation-interstitial atom impurity interactions contribute to strengthening. A number of mechanisms are possible, and these may have practical implications.

At high temperatures, $T/T_m > 0.4$, stable dispersions of fine particles are potent strengtheners of the Group VI-A. Small nonmetallic additions, i. e., dope, acting in solution or possibly at the grain boundaries may also contribute to the stability of non-sag tungsten filaments. Applied research in these areas is exploiting electron microscopy and other advanced methods, but is proceeding with little regard for mechanistic concepts of the underlying defect structure.

INTRODUCTION

Research in support of the development of a new material traditionally follows the empirical approach. Composition and processing, to the extent that these can be controlled, are correlated with structure and the properties of interest. In the past, such features of the structure as the crystal type, the lattice parameter, the grain size, the phase distribution, and the degree of cold work and anisotropy have been studied. These "classical" concepts of structure provide the base for the existing technology of the Group VI-A metals and their alloys now in use.

With the advent of electron microscopy and other new research tools, various features of the structure on the atomistic scale have recently become accessible. Progress in characterizing the nature of the electronic bonding is also being made. As a result, correlations with mechanical strength involving dislocations, cell structure, stacking faults, vacancies, and the character of the electronic bonding are actively being sought. The impact of advance structural concepts has been felt mainly by the basic research community; the technology of the Group VI-A metals so far seems to have benefited little. Today the practical metallurgist finds it difficult to cite a stronger Group VI-A alloy or heat-treated condition that was inspired by new knowledge of defect interactions or the electron bonding. There is growing awareness on both sides that a closer coupling of research and development efforts is desirable. Certainly, the rate of progress to be derived for the Group VI-A metals from consideration of the classical structure is likely to be much slower in the future. Therefore, it seems an appropriate time to consider the insights advanced concepts offer the technologist, as well as important practical problems that remain unresolved.

This report examines progress made in applying advanced techniques and structural concepts to the problem of strengthening the Group VI-A metals. It was prepared in recognition of the role that basic research is playing in the development of materials within the DMIC scope. It was presented, along with similar analyses, at a conference held by the National Physical Laboratory in Teddington, England. Since there is no basic difference between the mechanical behavior of molybdenum, chromium, and tungsten, they are discussed together. Three temperature ranges are considered. (1)* At low temperatures, $T/T_m < 0.2^{**}$, dislocations and grain boundaries play an important role in determining mechanical strength. At intermediate temperatures, $0.2 < T/T_m < 0.4$, dislocation-impurity atom interactions contribute to strength. Finally, at the higher temperatures, $T/T_m > 0.4$, stable dispersions of fine particles exert a potent strengthening effect.

*References are given on page 27.

** T_m is the melting point.

LOW-TEMPERATURE STRENGTH

Resistance to Flow

The low-temperature strength of the Group VI-A metals is important mainly because of the attending brittleness problem. As shown in Figure 1, the mechanical properties of unalloyed chromium, molybdenum, and tungsten bear a striking resemblance in the region $T/T_m < 0.2$. The yield strength is strongly temperature dependent in this range, and all three metals display brittle behavior at the lowest temperatures. (2, 3, 4) Conrad and Hayes (5, 6) associate the strong temperature dependence of the yield and flow strength of the Group VI-A metals with the process governing dislocation mobility, and treat this as a single, thermally activated process. Accordingly, the rate of dislocation movement at yielding is determined by: (1) the work of the effective shear stress τ^* (the difference between the applied stress τ and the long-range internal stress τ_μ which is proportional to the shear modulus) and (2) the intensity of thermal fluctuations. Figure 2 shows that the temperature dependence of the effective shear stress plotted against the relative temperature parameter $T-T_0/T_m$ (T_0 is the temperature at which τ becomes equal to τ_μ) is essentially the same for the Group VI-A metals. The activation energies for movement calculated in this way are independent of structure and purity, and, as illustrated in Figure 3, correlate with the shear modulus. These results support the idea that the process governing flow is the same in all three cases. Conrad and Hayes believe that the Peierls-Nabarro stress is the rate-controlling mechanism. However, other mechanisms such as the formation of vacancies at nonconservative jogs or cross slip cannot be ruled out. The important point is that the characteristic temperature dependence seems to be an intrinsic property of the Group VI-A metals.

Dislocation mobility has actually been measured in tungsten crystals by Schadler and Low (7). Their results, reproduced in Figure 4, tend to confirm the idea that the resistance to dislocation movement is enhanced at low temperatures. However, since the dislocation velocity at the yield stress is not the same at room temperature and 77 K, it must be concluded that the number of dislocations participating in flow, the process of nucleation, unpinning or multiplication, is also temperature dependent.

The measured stress dependence of dislocation velocity in tungsten (see Figure 4) is of the form $\bar{v} = (\frac{\tau}{\tau_0})^m$, where \bar{v} is the average velocity, τ the applied shear stress, τ_0 the stress corresponding to unit velocity, and m , an empirical parameter describing the stress dependence. Values of m for the group VI-A and other metals are summarized in Table 1. The fact that the Group VI-A metals are characterized by relatively small m values has several important implications:

(1) Yield Points. According to the dislocation dynamics treatments of yielding (11, 12), metals exhibiting small m values, e. g., $m < 20$, will display a large abrupt yield drop when the initial density of mobile dislocations is small, e. g., 10^3 - 10^4 cm/cm³ or less. In the case of Group VI-A grown-in dislocations can be strongly pinned by impurity atmospheres or precipitates. This is illustrated in Figure 5a. Thus these metals frequently exhibit pronounced yield drops. On the other

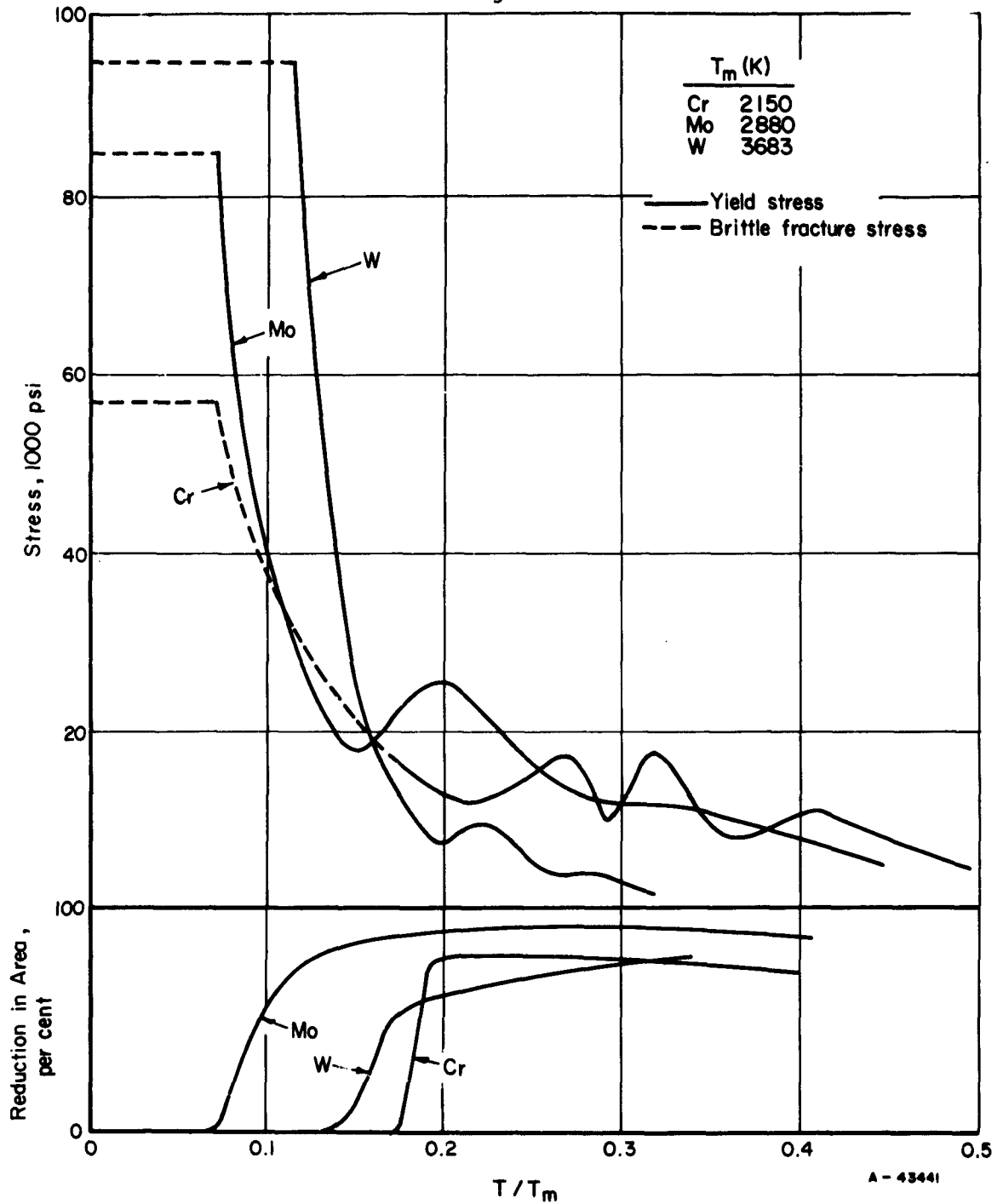


FIGURE 1. STRENGTH PROPERTIES OF THE GROUP VI-A METALS(2,3,4)
 PLOTTED AS A FUNCTION OF HOMOLOGOUS TEST
 TEMPERATURE

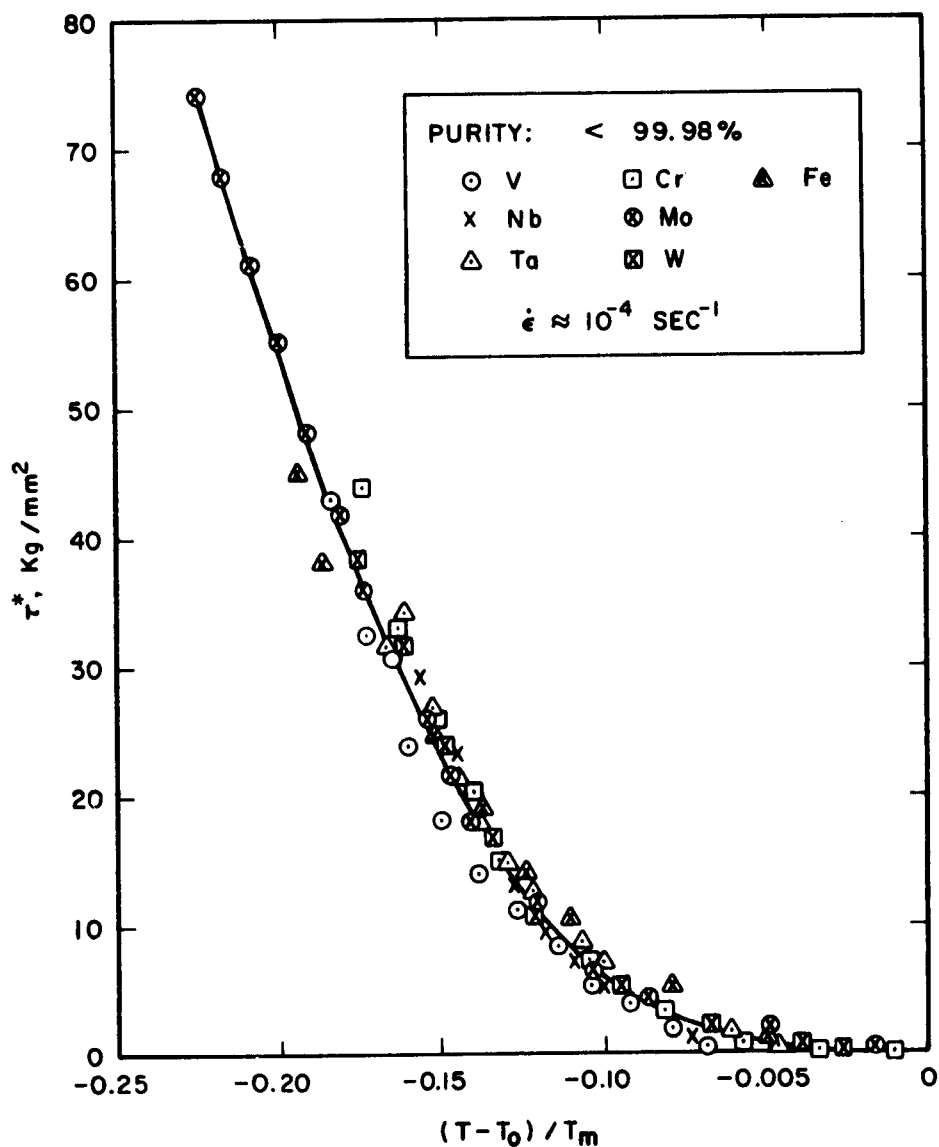


FIGURE 2. CORRELATION OF THE THERMAL COMPONENT OF THE YIELD STRESS τ^* WITH THE PARAMETER $\frac{(T - T_0)}{T_m}$ FOR THE B. C. C. TRANSITION METALS

After Conrad and Hayes(6).

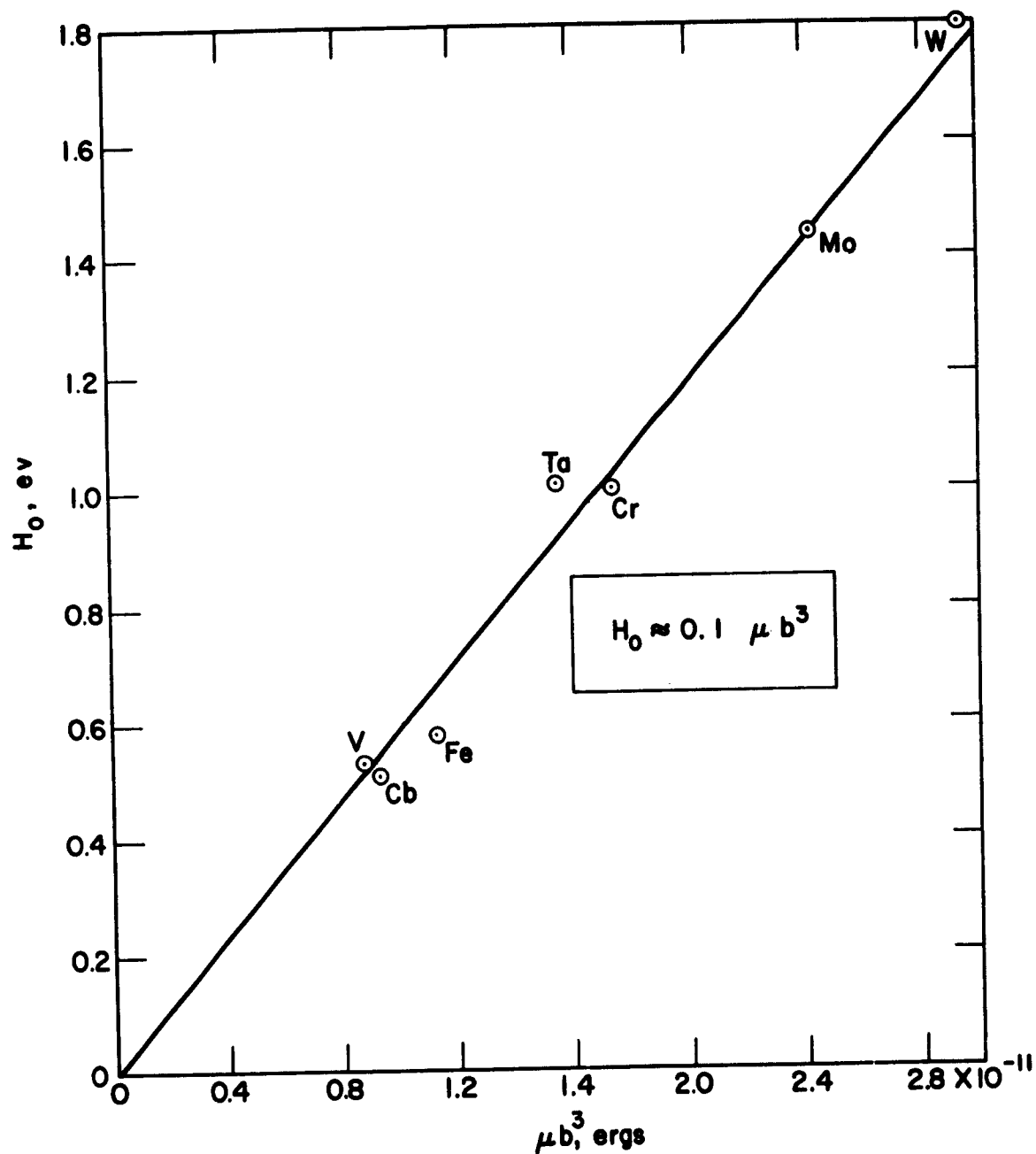


FIGURE 3. CORRELATION AMONG CALCULATED ACTIVATION ENERGY VALUES FOR YIELDING AND THE SHEAR MODULUS

After Conrad and Hayes(5).

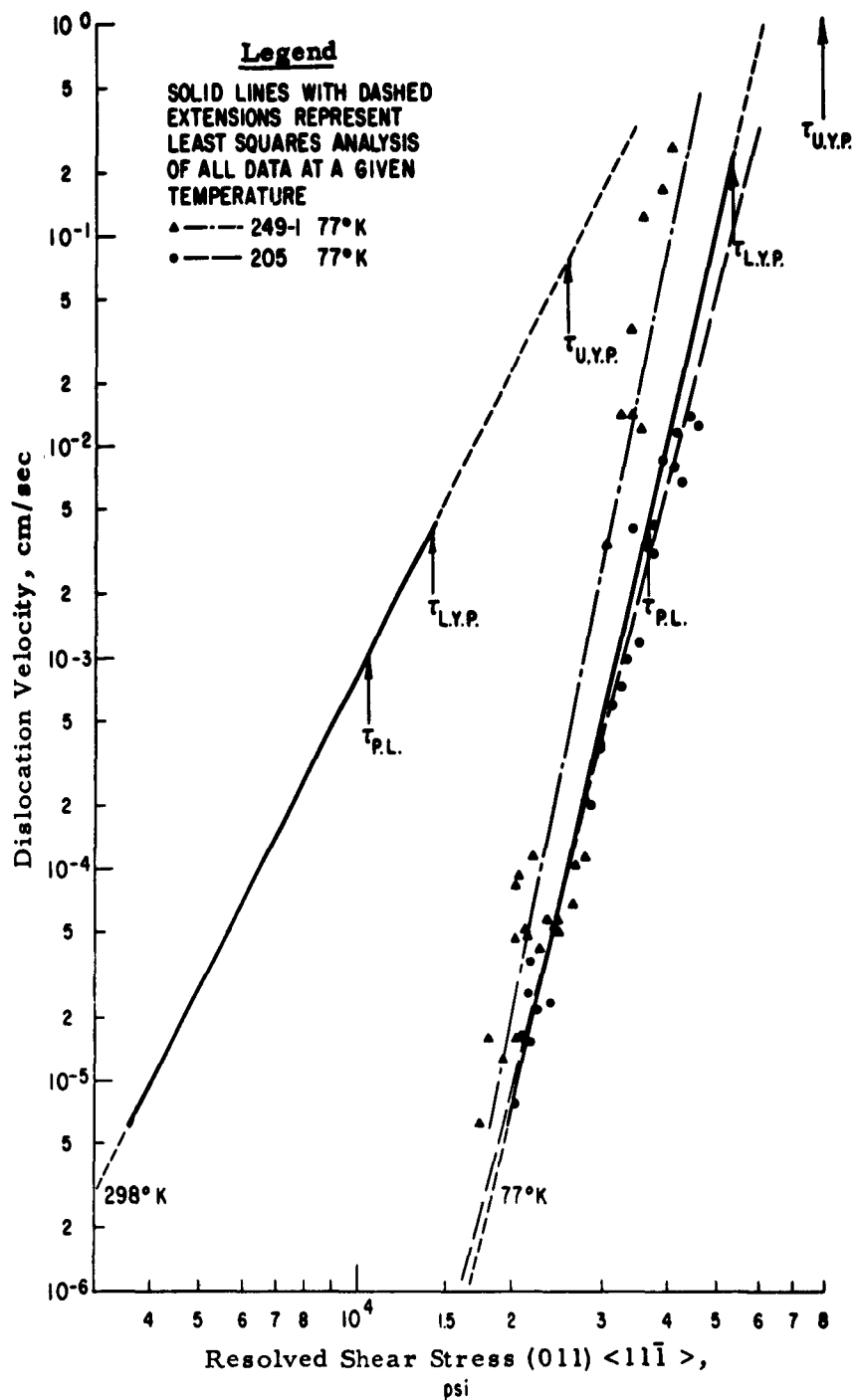


FIGURE 4. MOBILITY OF EDGE DISLOCATIONS MOVING ON (011) TYPE PLANES IN TUNGSTEN AT 77 and 298 K

After Schadler and Low(7).



b.

a.

FIGURE 5. TRANSMISSION ELECTRON MICROGRAPHS OF UNALLOYED MOLYBDENUM SHOWING EXAMPLES OF (a) GROWN-IN DISLOCATIONS DECORATED WITH PRECIPITATE PARTICLES AND (b) DISLOCATIONS GENERATED BY A PRECIPITATE PARTICLE, PROBABLY A CARBIDE

hand, Figure 5b illustrates that precipitate particles in the Group VI-A metals also are sites where free dislocation can be generated easily. Probably, for this reason, yield points are sometimes not observed even when the grown-in dislocations are locked.

(2) Strain-Rate Sensitivity. The dislocation dynamics treatment also shows that m is closely related to the strain-rate dependence of the yield and flow stress.⁽¹⁾ Thus, the Group VI-A metals, which have relatively small m values, are unusually strain-rate sensitive. At ordinary temperatures, the high rates of deformation encountered in forming operations thus lead to high stresses and promote brittle behavior.

TABLE 1. VALUES OF THE DISLOCATION VELOCITY PARAMETER, m , FOR DIFFERENT MATERIALS^(a)

Material	m	Reference
Si (600 - 900 C)	1.4	8
Ge (420 - 700 C)	1.4-1.9	8
Cr	$\sim 7^{(b)}$	9
Mo	$\sim 8^{(b)}$	9
W	5	7
LiF	14.5	12
Fe-3.25 Si	35	10
Cu	$\sim 200^{(b)}$	11
Ag	$\sim 300^{(b)}$	11

(a) At room temperature unless otherwise noted.

(b) Estimated from the strain-rate dependence of yield stress.

Resistance to Fracture

Hahn, Gilbert, and Reid⁽¹³⁾ have recently gone one step further in this direction. Using the simple model of a crack illustrated in Figure 6 and the dislocation dynamics treatment of yielding, they have calculated the stresses and extent of yielding in advance of a moving crack. Results of these calculations, summarized in Figure 7, show that for m values characteristic of Group VI-A, e. g., $m = 5$, stresses of the order of the theoretical strength are generated at the tip of fast-moving cracks.* Thus the propensity for cleavage also seems to be an intrinsic feature of the Group VI-A metals.

The low-temperature brittleness tendency of the Group VI-A metals is accentuated by weak grain boundaries. Fractographic studies of brittle samples invariably reveal a grain-boundary rupture at the fracture origin^(14, 15, 16), as illustrated in Figure 8. Such a rupture is believed to be the source of a fast-moving crack capable of penetrating adjoining grains by transcrystalline cleavage.

*The calculations illustrated are based on an assumed rate of dislocation multiplication $C = 10^9$ cm per cm³ per per cent strain and an initial density of mobile dislocations $\rho_0 = 10^3$ cm per cm³.

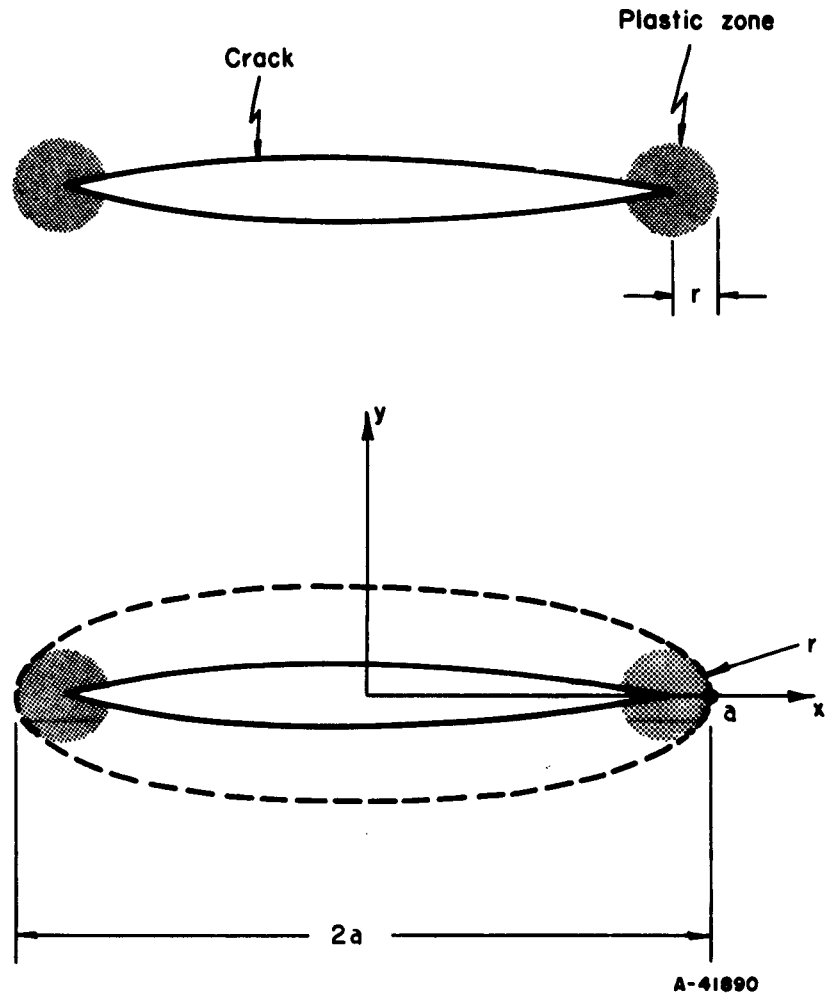


FIGURE 6. MODEL OF A PARTIALLY RELAXED CRACK AFTER HAHN, GILBERT, AND REID(13)

It is assumed that (1) the plastic zone (shaded) is circular, (2) the elastic-plastic boundary corresponds to a strain of 0.1 per cent, (3) stresses within the zone are substantially relaxed, and (4) a plane stress condition prevails. The crack-plastic zone complex (dashed lines) can then be treated as an elliptical hole.

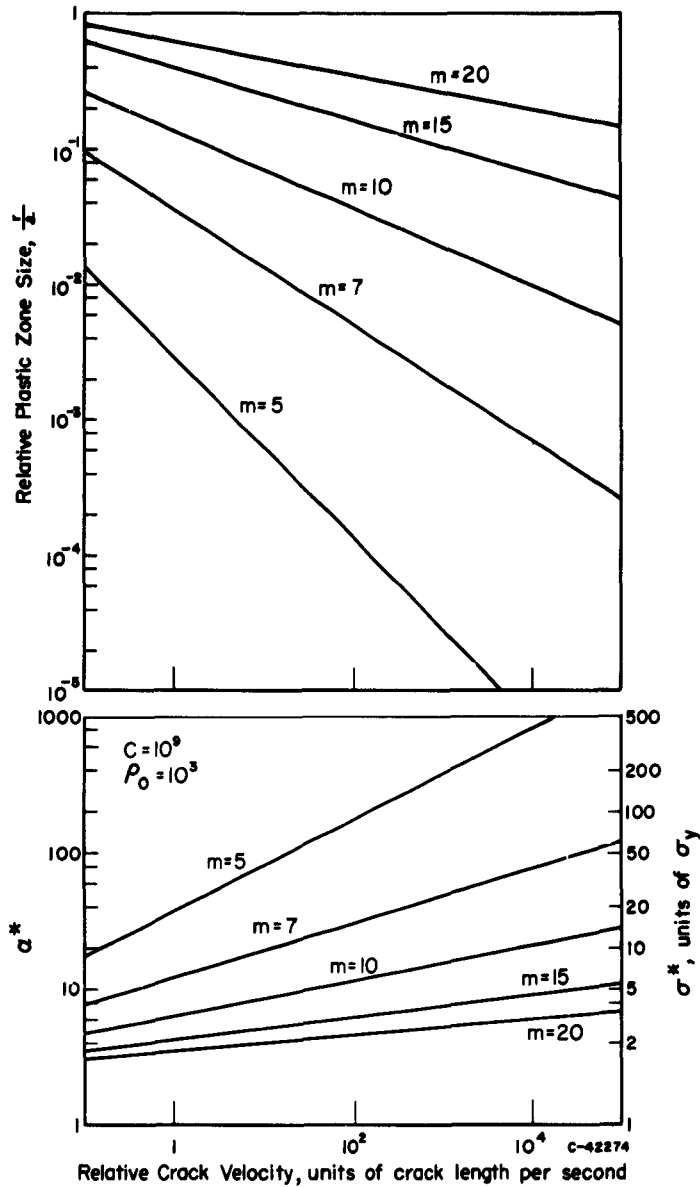
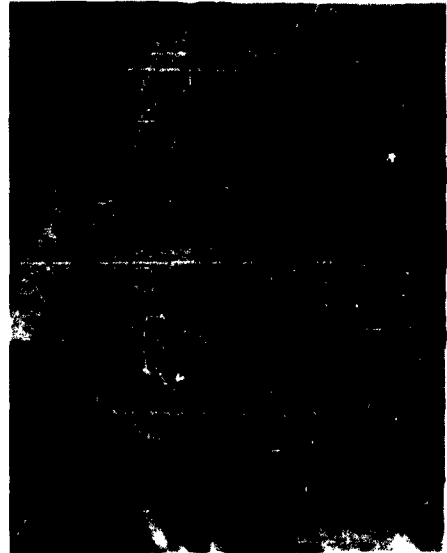


FIGURE 7. CALCULATED VALUES OF STRESS AND PLASTIC ZONE SIZE AHEAD OF A MOVING CRACK, AFTER HAHN, GILBERT AND REID(13)

The relative plastic zone size, $\frac{r}{a}$ (top diagram) and the maximum tensile stress, σ^* , and stress concentration, α^* , (bottom diagram) are plotted as a function of crack velocity for various values of the parameter m .



2000X a. Chromium



750X b. Molybdenum



1000X c. Tungsten

FIGURE 8. FRACTOGRAPHS SHOWING EXAMPLES OF GRAIN-BOUNDARY RUPTURES AT THE ORIGIN OF BRITTLE CLEAVAGE FRACTURES OF CHROMIUM⁽¹⁴⁾, MOLYBDENUM⁽¹⁵⁾, AND TUNGSTEN⁽¹⁶⁾ TEST BARS

The grain-boundary weakness is related to the low solubility for interstitials displayed by chromium, molybdenum, and tungsten, and the tendency for these atoms to segregate and precipitate at the grain boundaries. Furthermore, the high surface tension of the Group VI-A metals promotes wetting of insoluble interstitial phases and thus enhances the embrittling effect of the interstitial impurities. In the absence of grain boundaries, single crystals are ductile to much lower temperatures and only become brittle with the onset of deformation twinning. A high degree of purification can also restore ductility. This has recently been demonstrated by Lawley, Van der Syde, and Maddin⁽¹⁷⁾ who show that multiple-zone-passed molybdenum is ductile at 4 K even after it is recrystallized. However, neither single crystals nor purification are practical answers.

Severe working to break down the grain-boundary structure improves ductility, but this cannot always be accomplished, because the grain-boundary weakness and intergranular cracking are encountered over the entire temperature range to the melting point. Hot brittleness was solved for tungsten and molybdenum some 50 years ago by Fink⁽¹⁸⁾ working at the General Electric Company. He did this by consolidating pure tungsten powder at high temperatures achieved by internal-resistance heating and then swaging (to avoid secondary tensile cracking) at progressively lower temperatures (to avoid recrystallization).

The technological requirement for larger ingots has led to consolidation by arc melting. Ingots prepared in this way have a characteristic large-grained columnar structure. This is comprised of long crystal fingers, extending radially upward from a centrally equiaxed core, a structure not unlike that of chill castings. The hot working of such large-grained ingots is much more difficult than that of the fine-grained sintered powder ingots, and they are susceptible to intergranular cracking during the breakdown forging. In the case of molybdenum, the cracking was combatted by deoxidation with carbon. In this way deleterious patches of Mo-MoO₂ eutectic at the grain boundaries are replaced by more desirable non-wetting Mo₂C carbide platelets. Structural molybdenum alloys are even more difficult to hot forge, and a process was developed for breaking up the coarse, as-cast structure by extrusion, a deformation process in which secondary tensile cracking is controlled by lateral support of the extrusion die walls. However, in some of the newer high-strength alloys severe cracking is now being encountered after ingot solidification and cooling before this condition can be relieved by hot working.

To summarize, the brittleness of the Group VI-A metals has as its origin the intrinsic low mobility of the dislocations and the weakness of the grain boundaries. Both of these factors deserve more attention. A better understanding of the structure of grain boundaries, their binding energies, and the separate effects of segregates and precipitates might provide better answers to the grain-boundary cracking problem than the mechanical solutions currently employed. Studies of rhenium additions described in the next section offer some hope that the brittleness tendency can be inhibited by alloying.

The Rhenium Alloying Effect

In 1955 Geach and Hughes⁽¹⁹⁾ reported the discovery that both the workability and low-temperature ductility of molybdenum and tungsten are greatly improved by replacing about one-third of the Group VI-A metal atoms by rhenium. This effect has since been demonstrated for chromium⁽²⁰⁾; typical results for the Group VI-A metals are illustrated in Figure 9. More recently, results of Pugh, Amra, and Hurd⁽²³⁾, summarized in Figure 10, have brought to light that smaller rhenium additions in the range 1-5 per cent

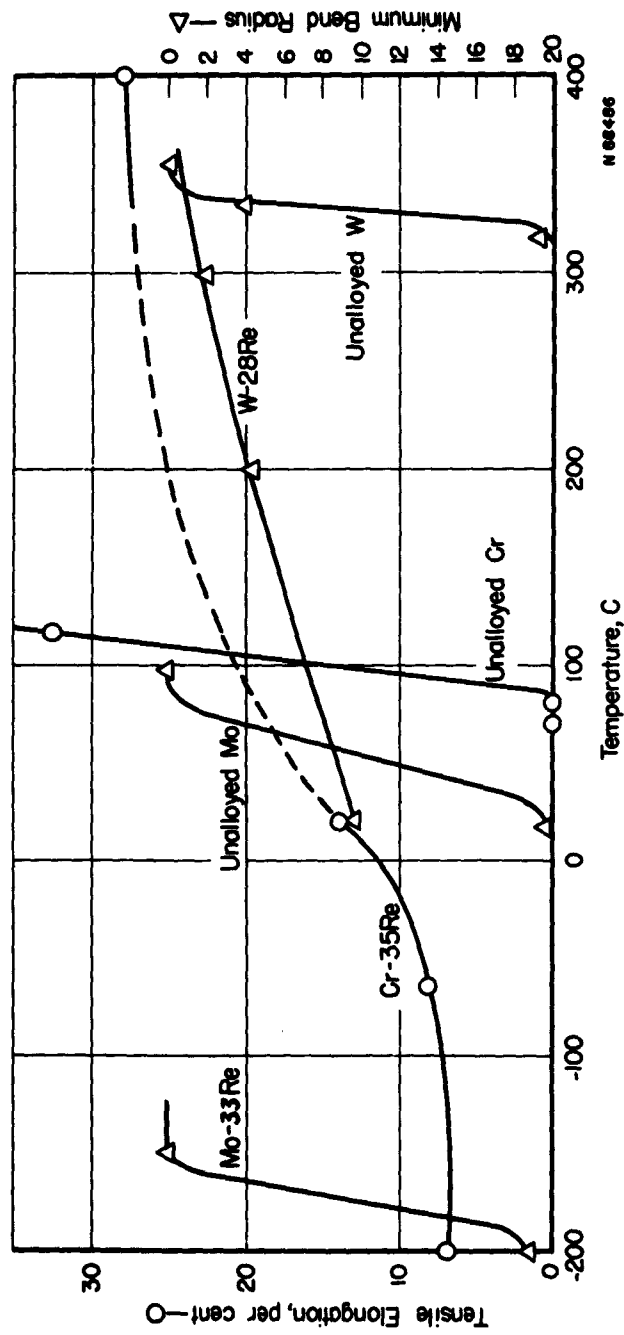


FIGURE 9. EFFECT OF HIGH RHENIUM ADDITIONS (ATOMIC PER CENT) ON DUCTILE-BRITTLE TRANSITION IN CHROMIUM(20), MOLYBDENUM(21,22) AND TUNGSTEN(21,22)

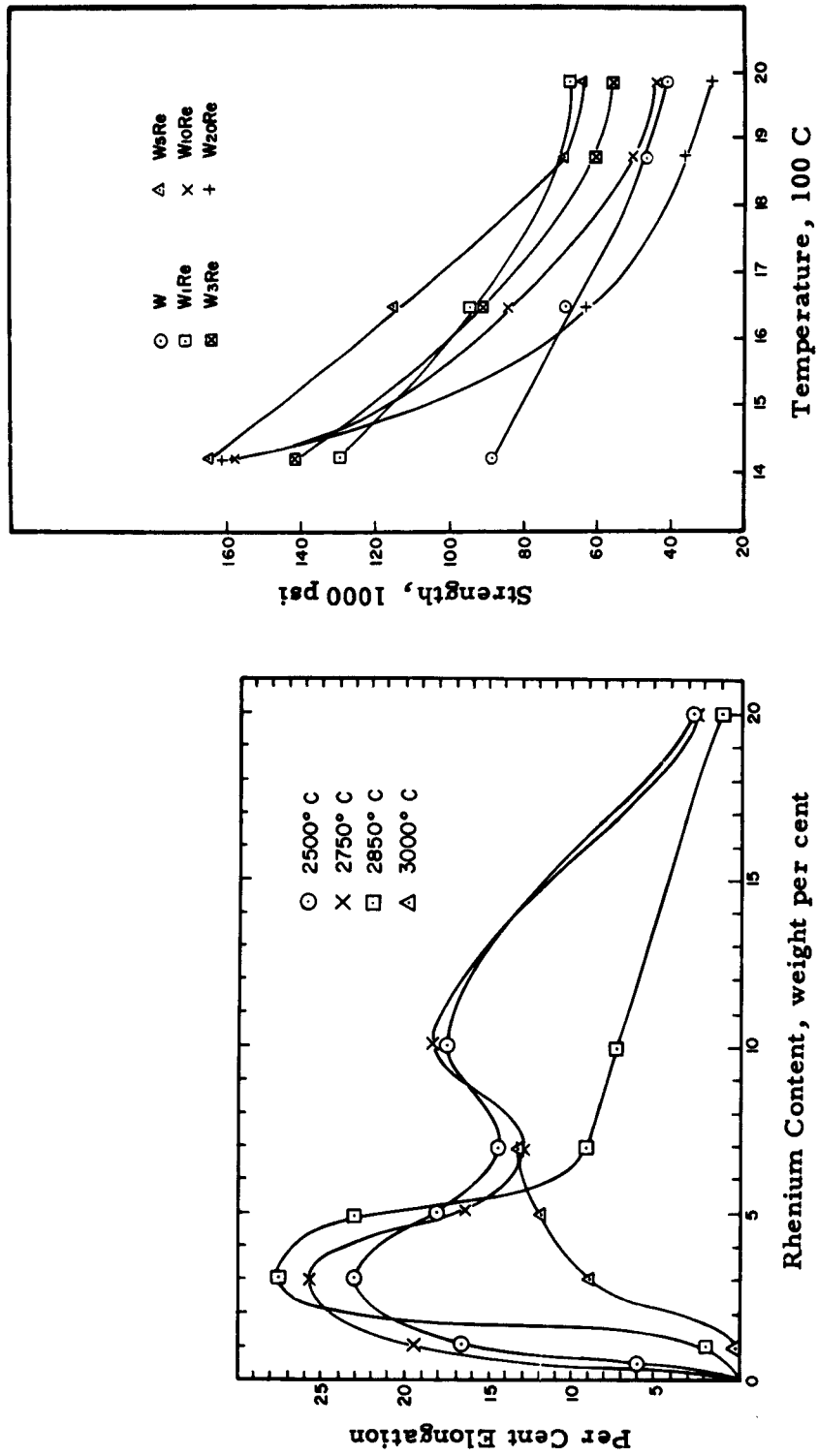


FIGURE 10. EFFECT OF RHENIUM ADDITIONS TO 218 DOPED TUNGSTEN WIRE ON:
 (a) ROOM TEMPERATURE DUCTILITY AFTER HEATING 3 MINUTES AT
 THE INDICATED TEMPERATURES, AND (b) ELEVATED TEMPERATURE
 STRENGTH

After Pugh, Amra, and Hurd(23).

accomplish a large improvement both in the room-temperature ductility and high-temperature strength of tungsten wire which is "doped" to recrystallize with the large-grained elongated non-sag structure.

There is evidence that rhenium influences dissolved interstitial atoms as well as the properties of the grain boundary and the lattice, but the origin of the ductilizing effect remains elusive. Small additions of rhenium tend to decrease hardness, with a minimum occurring at about 5 atomic per cent. This effect, illustrated in Figure 11, has been interpreted as a reduction in interstitial solubility similar to the effect of Group VI-A, VII-A, and VIII-A additions on oxygen solubility in columbium⁽²⁵⁾. Since hardness minima produced by Group IV-A, V-A, and VI-A additions are also observed (see Figure 11), the reduction in interstitial solubility may not be a factor in the ductilizing mechanism. Higher rhenium contents up to 20 atomic per cent strengthen the Group VI-A metals markedly at low temperatures. Beyond 20 per cent to the solubility limit (30-40 atomic per cent), strength remains relatively unchanged. In this range enhanced ductility is accompanied by increasing amounts of mechanical twinning. It has been suggested that twinning is a result of a decrease in the stacking-fault energy. As shown in Figure 12, dislocation segments in a Mo-35Re alloy are straighter and do appear to be confined to individual slip planes, but stacking faults from extended dislocations so far have not been observed. Fractographic studies reveal that fracture of rhenium alloys is predominantly intergranular and suggest that crack propagation by cleavage is more difficult. The alternative of twinning may offer these alloys a more responsive mechanism of stress relaxation, but the fact that the twins themselves do not produce cracks suggests that dislocation mobility is also affected. Another effect of rhenium, noted by Jaffee and co-workers⁽²⁶⁾, is the reduced wetting of grain boundaries by oxide phases. This indicates a lowering of the grain-boundary energy, along with the postulated decrease in twin-boundary and stacking-fault energy.

There is evidence that the rhenium-alloying effect is associated with an increased electron-to-atom ratio. If the bonding electrons in the Group VI-A metals are taken as 6 (s and d) electrons, the alloys showing maximum ductility and maximum twinning have 6.4 electrons per atom. Addition of all other transition metals from Groups VII-A and VIII-A, B, and C, which possess seven to ten (s and d) electrons in bonding orbitals, also produce increased mechanical twinning and a minimum in hardness. There are 12 such elements and 36 alloying systems which involve this behavior with the Group VI-A metals. In the range 6 to 6.4 electrons per atom, the electronic structure undergoes important changes. The electronic specific heat rises to a peak as illustrated in Figure 13. The critical temperatures for superconductivity also exhibit maxima at about the same electron-to-atom ratio. In both cases, this is associated with a peak in the density of states at the Fermi level. The increased tendency for twinning and the density of states may also be connected. Effects of rhenium thus have important practical and theoretical implications. Efforts are being made at Battelle to reproduce the beneficial effects of rhenium with more abundant alloying constituents. Further experimentation in this area also may provide insight into the role of the electronic bond in mechanical behavior.

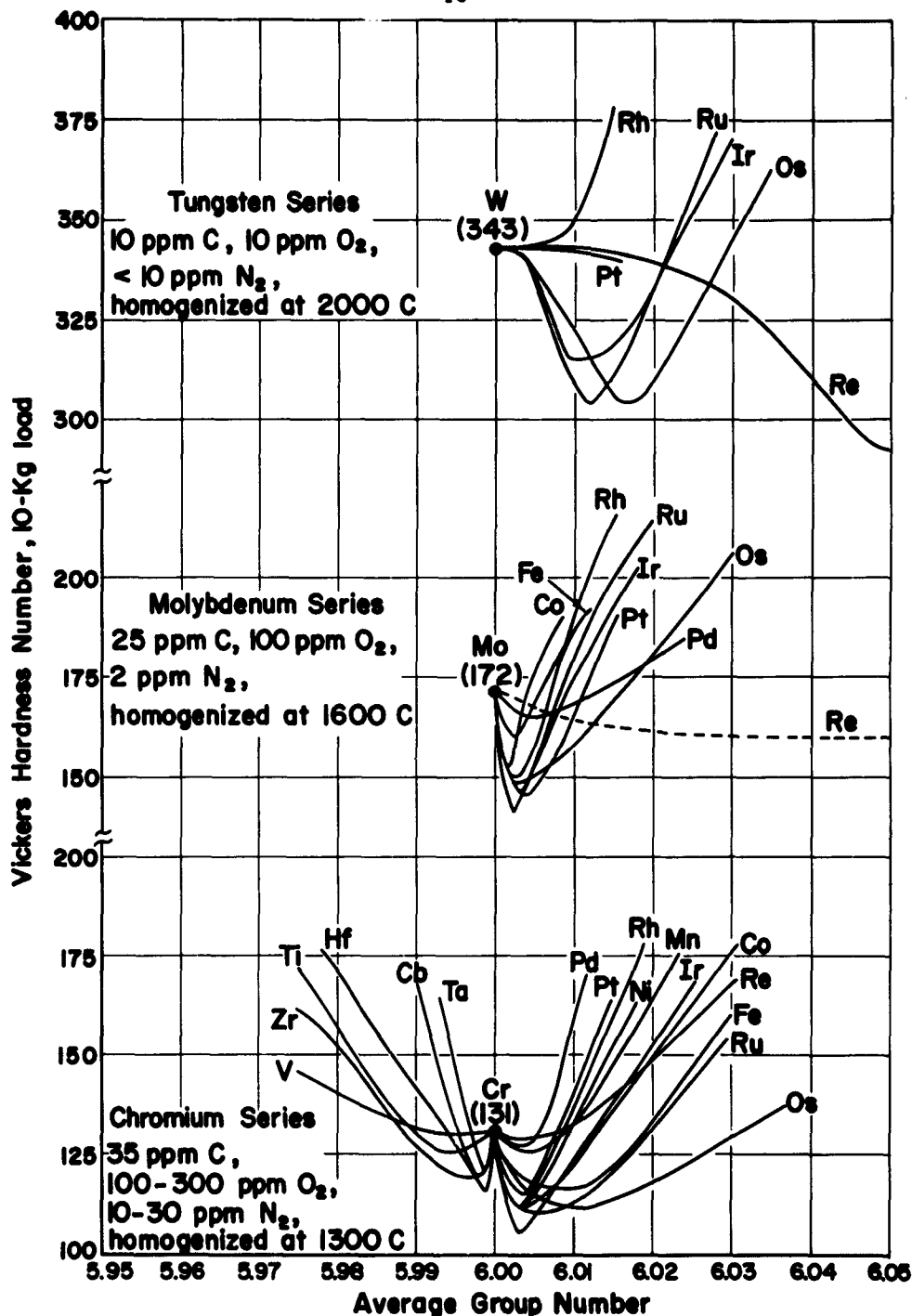


FIGURE 11. INFLUENCE OF ALLOYING WITH GROUP IV-A, V-A, VII-A AND VIII-A METAL ADDITIONS ON THE HARDNESS OF CHROMIUM, MOLYBDENUM AND TUNGSTEN AFTER ALLEN AND JAFFEE⁽²⁴⁾. ALLOY CONTENT IS DESCRIBED IN TERMS OF THE AVERAGE GROUP NUMBER

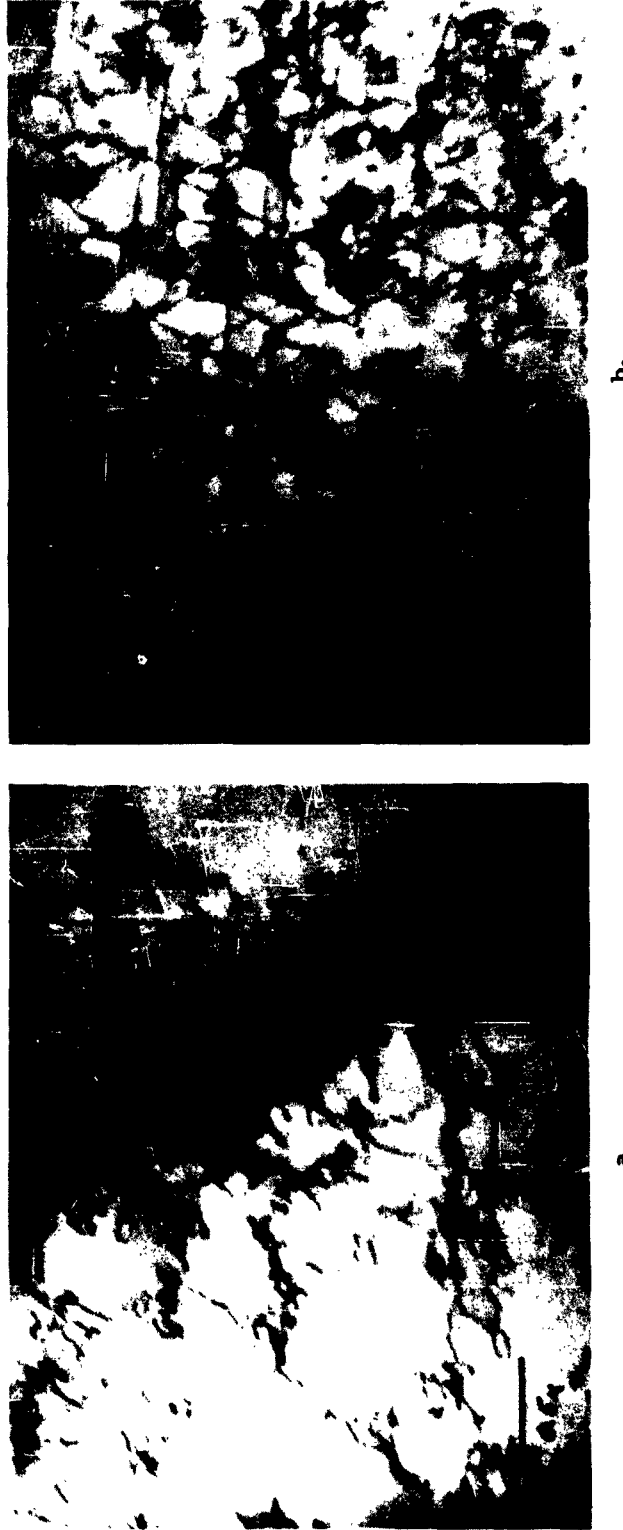


FIGURE 12. TRANSMISSION ELECTRON MICROGRAPHS OF LIGHTLY DEFORMED: (a) UNALLOYED MOLYBDENUM AND (b) THE Mo-35 Re ALLOY

The dislocations in the alloy are noticeably straighter and tend to be confined to slip planes.

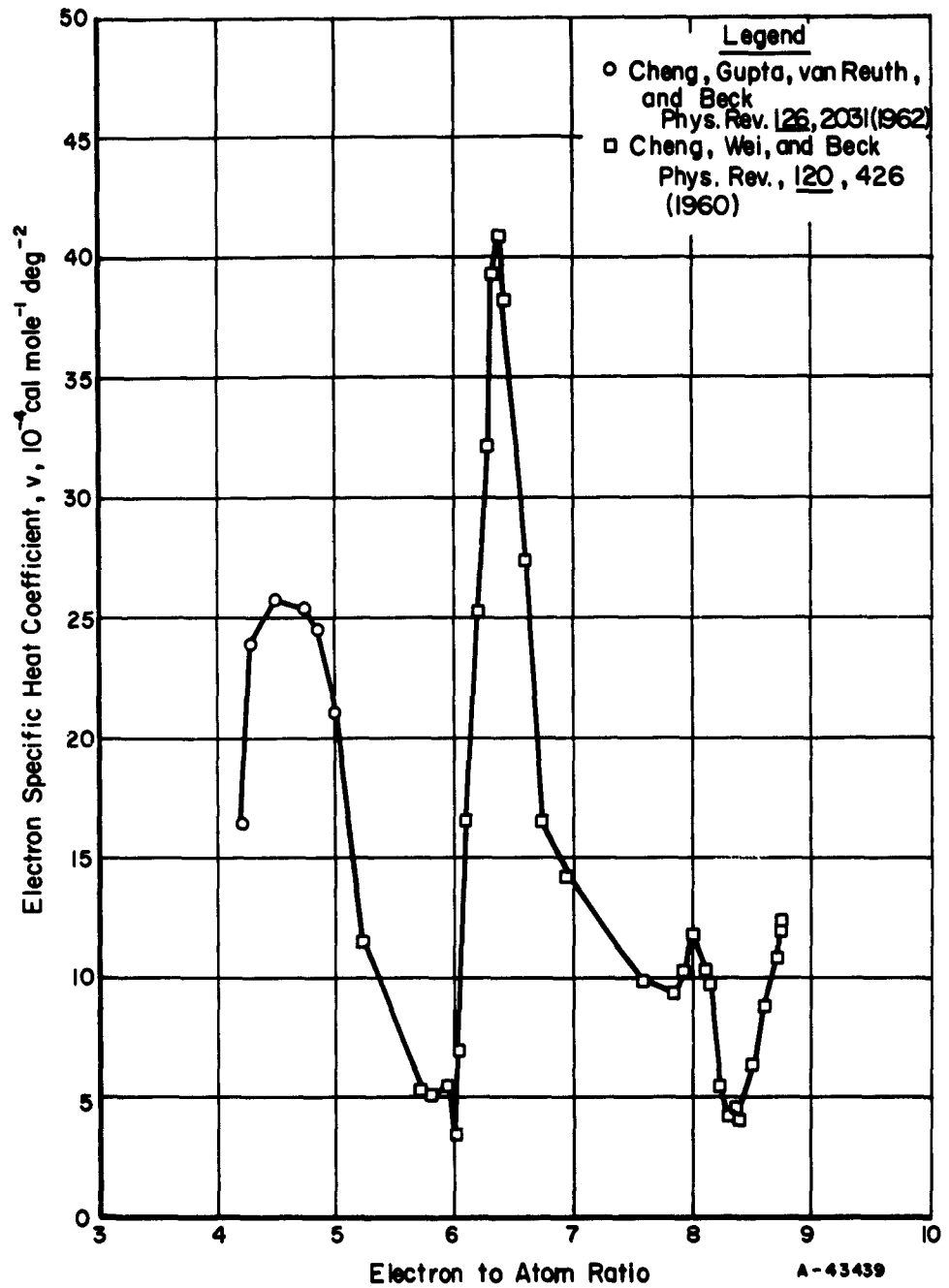


FIGURE 13. ELECTRONIC SPECIFIC HEAT OF SOME FIRST LONG-PERIOD BCC ALLOYS (Ti-V, V-Cr, Cr-Fe, AND Fe-Co)

INTERMEDIATE TEMPERATURES

The source of the strength of the Group VI-A metals at intermediate temperatures, $0.2 < T/T_m < 0.4$, has not been of practical interest. In this range adequate strength is obtained from alloying and processing designed to confer needed strength at the highest temperatures. However this region has received some attention, and features of strengthening peculiar to the unalloyed metals may have practical value.

As shown in Figure 14, temperature intervals are observed in this range within which the yield and flow strength is markedly enhanced. The regions of peak strength, which may overlap, are accompanied by:

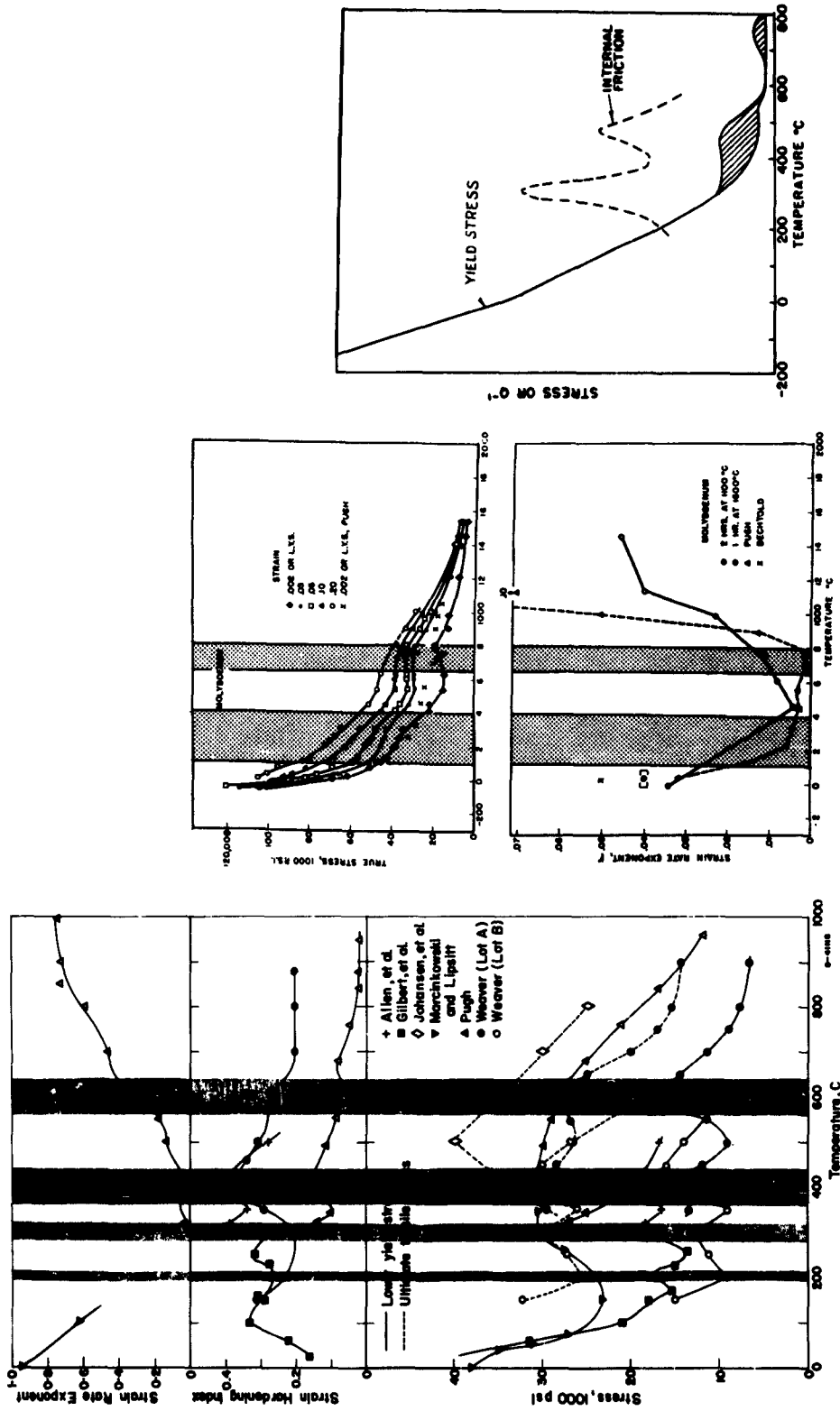
- (1) Diminished strain-rate sensitivity,
- (2) Increases in the rate of work hardening, and frequently by
- (3) Some loss in ductility.

Such peaks are also displayed by iron and the metals of Group V-A, and it is likely that similar mechanisms are at play. Several explanations involving dislocation-interstitial atom interactions have been advanced. Originally the strength increase was attributed to strain-induced precipitation, a strain-aging process occurring simultaneously with deformation. Cottrell⁽³⁵⁾ suggested that interstitial atoms are sufficiently mobile at these temperatures to form atmospheres about moving dislocations, thus acting as a drag. More recently Schoeck and Seeger⁽³⁶⁾ have proposed stress-induced ordering of interstitials (the Snoek effect) in the strain field of the moving dislocation as the source of the drag. According to these theories, the strengthening will be confined to a narrow temperature range determined by the mobility of the interstitial atom. Since these mechanisms can apply to different interstitial species present, and also to interstitial atom-vacancy or interstitial-substitutional impurity complexes, they are consistent with the multiplicity of strength peaks observed. Schnitzel⁽³⁴⁾ has recently correlated two Snoek internal-friction peaks observed in tungsten with the 475 and 750 C strength peak displayed by this material (see Figure 14). This work lends weight to the Schoeck and Seeger argument, but does not rule out the other mechanisms.

The results of Chang⁽³⁷⁾, reproduced in Figure 15, suggest this family of strengthening phenomena can have practical implications. The results for the commercial molybdenum Mo-TZC alloy in an annealed condition display an analogous strength peak in the vicinity of 2400 F (1300 C). Thus it may be possible to exploit dislocation interactions of this type at the highest temperatures. These phenomena can also be useful for studying indirectly the mobility of interstitial atoms, the velocity of dislocations, and the strength of interstitial-substitutional atom interactions in solid solution.

HIGH-TEMPERATURE STRENGTH

Of the considerable effort to develop Group VI-A alloys with superior strength at high temperatures, $T/T_m > 0.4$, which are not brittle and can be processed similar to



a. Chromium
b. Molybdenum
c. Tungsten

FIGURE 14. THE STRENGTH OF UNALLOYED (a) CHROMIUM(3,27-31), (b) MOLYBDENUM(32,33), AND (c) TUNGSTEN(34) AT INTERMEDIATE TEMPERATURES

The test data display regions of enhanced strength associated with interstitial impurities.

INTERMEDIATE TEMPERATURES

The source of the strength of the Group VI-A metals at intermediate temperatures, $0.2 < T/T_m < 0.4$, has not been of practical interest. In this range adequate strength is obtained from alloying and processing designed to confer needed strength at the highest temperatures. However this region has received some attention, and features of strengthening peculiar to the unalloyed metals may have practical value.

As shown in Figure 14, temperature intervals are observed in this range within which the yield and flow strength is markedly enhanced. The regions of peak strength, which may overlap, are accompanied by:

- (1) Diminished strain-rate sensitivity,
- (2) Increases in the rate of work hardening, and frequently by
- (3) Some loss in ductility.

Such peaks are also displayed by iron and the metals of Group V-A, and it is likely that similar mechanisms are at play. Several explanations involving dislocation-interstitial atom interactions have been advanced. Originally the strength increase was attributed to strain-induced precipitation, a strain-aging process occurring simultaneously with deformation. Cottrell⁽³⁵⁾ suggested that interstitial atoms are sufficiently mobile at these temperatures to form atmospheres about moving dislocations, thus acting as a drag. More recently Schoeck and Seeger⁽³⁶⁾ have proposed stress-induced ordering of interstitials (the Snoek effect) in the strain field of the moving dislocation as the source of the drag. According to these theories, the strengthening will be confined to a narrow temperature range determined by the mobility of the interstitial atom. Since these mechanisms can apply to different interstitial species present, and also to interstitial atom-vacancy or interstitial-substitutional impurity complexes, they are consistent with the multiplicity of strength peaks observed. Schnitzel⁽³⁴⁾ has recently correlated two Snoek internal-friction peaks observed in tungsten with the 475 and 750 C strength peak displayed by this material (see Figure 14). This work lends weight to the Schoeck and Seeger argument, but does not rule out the other mechanisms.

The results of Chang⁽³⁷⁾, reproduced in Figure 15, suggest this family of strengthening phenomena can have practical implications. The results for the commercial molybdenum Mo-TZC alloy in an annealed condition display an analogous strength peak in the vicinity of 2400 F (1300 C). Thus it may be possible to exploit dislocation interactions of this type at the highest temperatures. These phenomena can also be useful for studying indirectly the mobility of interstitial atoms, the velocity of dislocations, and the strength of interstitial-substitutional atom interactions in solid solution.

HIGH-TEMPERATURE STRENGTH

Of the considerable effort to develop Group VI-A alloys with superior strength at high temperatures, $T/T_m > 0.4$, which are not brittle and can be processed similar to

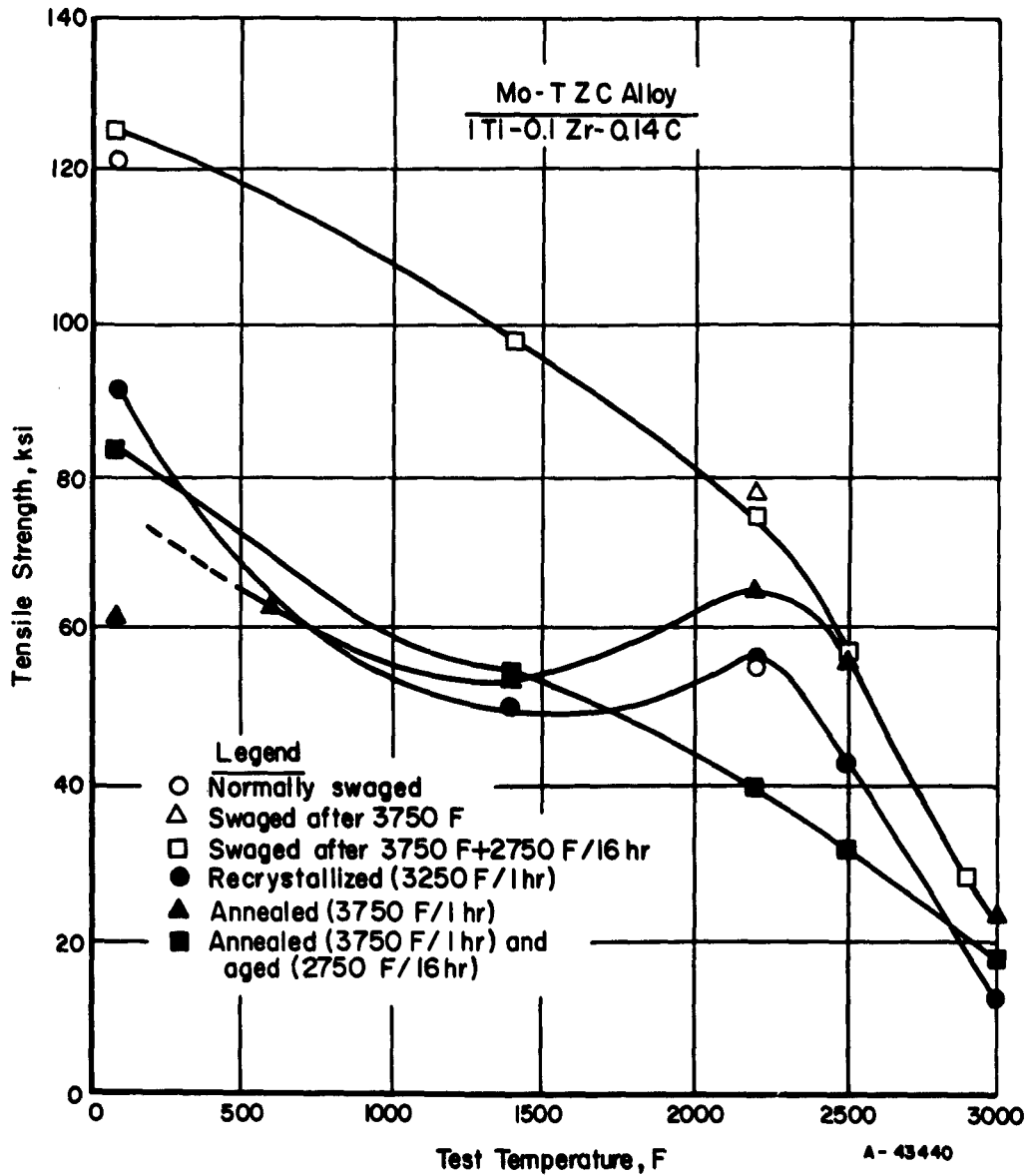


FIGURE 15. EFFECT OF TEMPERATURE ON THE TENSILE PROPERTIES OF NON-WORKED Mo-TZC ALLOY AFTER CHANGE⁽³⁷⁾

the unalloyed metals, the strengthening afforded by fine-particle dispersions has proven most successful. In the case of molybdenum, the development can be traced to alloy studies at The Climax Molybdenum Company, described by Semchyshen⁽³⁸⁾. This work demonstrated potent strengthening effects for 0.1-1.0 per cent additions of reactive metals like titanium and zirconium. The reason for the effectiveness of such small concentrations was not understood until it was discovered the optimum amount of titanium corresponded with a critical ratio of titanium to the residual carbon content left after deoxidation. When the carbon content was increased, maximum strengthening was found at higher titanium contents; thus the strengthening could be attributed to fine submicroscopic titanium carbide particles, rather than a solid-solution effect.

Further progress along these lines has been made by Chang^(37, 39), who has made extensive studies of the structure and properties of molybdenum alloys containing titanium, zirconium, and carbon. By use of both electron-microscopy and X-ray techniques the stability ranges of the various carbides, shown in Figure 16, were delineated and the precipitation of TiC identified as the principal hardening reaction. By employing proper solution and aging treatments followed by warm working, fine dispersions exhibiting high strength and room-temperature ductility have been developed. As an illustration, strength properties of the molybdenum-TZC alloy shown in Figure 15 may be compared with those of the unalloyed material (see Figure 13b). Recent work has also shown that these same alloys, when produced by powder-metallurgy techniques, can attain the same high strength level possible in the melted alloys. The reaction between the reactive metal addition and the adsorbed oxygen on the surface of the metal powder must be avoided during processing, and solution treatment and aging to precipitate the fine carbide is necessary. Thus, early speculation that the exceptional high-temperature strength of the arc-cast molybdenum-base alloy was the result of a submicroscopic, coherent carbide precipitated at high temperature during the solidification process, has been proven to be incorrect. Rather classical precipitation-hardening processes prevail for both arc-cast and powder-metallurgy alloys.

If insoluble dispersions are distinguished from precipitates from solid solution, it appears that only the oxides are stable in Group VI-A metals. Early work on oxide dispersions in powder-metallurgy molybdenum by Jaffee and co-workers⁽⁴⁰⁾ indicated that improvement in the high-temperature strength could be achieved by a fine, stable dispersion of the oxides of the reactive metals titanium and zirconium in approximately the same metallic concentrations as are effective in the arc-melted alloys described by Semchyshen. However, the magnitude of the increase in elevated-temperature strength with the oxides is less than that developed by carbide precipitates.

The development of precipitation-hardened, tungsten-base alloys has not followed the molybdenum history, chiefly because of different requirements. The dispersed phase commonly used in tungsten is thorium oxide, which thermodynamically is the most stable oxide. Thorium increases electron emission of tungsten and reduces grain growth. Thoriated tungsten wire is a common product in the lamp industry, but only recently has it been produced in other forms. Up to 8 volume per cent ThO₂ will reduce the ductile-to-brittle transition temperature to about the same or somewhat greater extent than would be expected from the reduced grain size⁽⁴¹⁾. Thorium also increases the elevated-temperature strength⁽⁴²⁾.

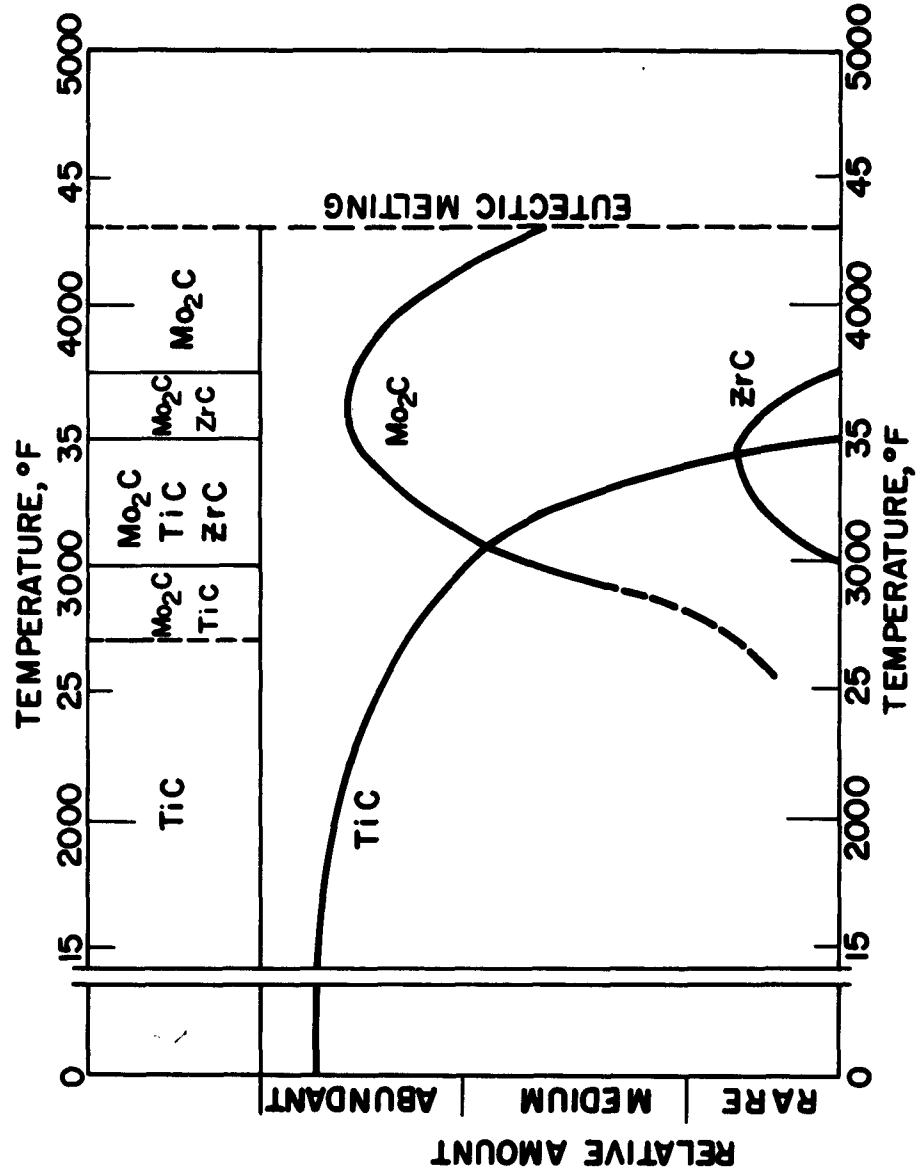


FIGURE 16. APPROXIMATE CARBIDE STABILITY RANGES IN Mo-Ti-Zr
(Mo-1Ti, -0.1 Zr-0.14 C) ALLOY
After Chang and Perlmuter(39).

Improvement in high-temperature properties of materials, through dispersion of a second phase, has been the subject of considerable research, and a number of models based on dislocation interactions have been proposed⁽⁴³⁾. However, the development of precipitation-hardened and dispersion-hardened Group VI-A alloys has not benefited from the theoretical treatments. The qualitative principles laid down by Jeffries forty years ago, with minor modification, still guide the development of dispersed-phase strengthened Group VI-A materials. These are: the dispersed particles should be as fine as possible, with minimum mean free matrix path, possess high chemical and size stability, with minimum solubility in the matrix, and high modulus of elasticity. Basic questions involving the bonding (coherency) between the dispersed phase and the matrix, interfacial energy, interaction between dislocations and the dispersed phases, etc., have not yet been treated nor applied to the high-temperature Group VI-A development effect.

The development of non-sag tungsten filaments is an example of the strengthening that has been achieved at the highest temperatures. Tungsten wire with a very stable large-grained, elongated structure was discovered accidentally when tungstic oxide was heated in Battersea crucibles, and picked up approximately 0.1% Al_2O_3 and SiO_2 . After processing to tungsten wire and recrystallizing at high temperature, the stable elongated-grained structure was developed. This, the Battersea process, has led to more direct methods of adding the nonmetallic addition known as "dope". The dope generally is a potassium-aluminum silicate addition and is added to the tungstic oxide or ammonium para-tungstate before reduction. Most of the addition evaporates during processing; a small amount remains as an insoluble inclusion, and a still smaller amount is believed to dissolve in the metal. The mechanism involved in developing the non-sag structure is still uncertain. There are two main viewpoints. One theory proffered by Milner⁽⁴⁴⁾ and Mannerkoski⁽⁴⁵⁾ argues that the chief role played by the dope results from the minute amount of solute in solid solution, which acts as Cottrell drag atmospheres on the dislocations. Mannerkoski proposes that the movement of screw dislocations during annealing occurs along the wire direction, whereas the edge dislocations move perpendicular to the wire axis. The atmospheres are thought to be unable to move with the screw dislocations, while they move with and drag the edges. Thus, grain-boundary motion is not impeded in the wire axis, and a large, elongated, recrystallized grain size results. The alternative viewpoint on non-sag tungsten is that of Reick⁽⁴⁶⁾, who proposed a "damaged dope skin" around the grain boundaries, which are fibered in the working process. During recrystallization, two preferred orientations result, [531] and [110]. The [531] oriented grains apparently can penetrate the damaged areas in the dope skin, while the [110] oriented grains cannot. Elongated grains with a [531] wire texture result. At this stage, these mechanisms are proposed without sufficient experimental evidence to permit a choice.

CONCLUDING REMARKS

The Group VI-A metals and alloys now in use owe their development to classical concepts of structure. The conclusion to be drawn from this review is that the more advanced concepts of structure are not yet being exploited. To be sure, such work as the studies of dislocations has been very useful in rationalizing the source of strength and brittleness of the Group VI-A metals at low temperatures. At the same time it has yielded little in the way of practical solutions to the brittleness problem. Work to

improve high-temperature strength, today, is proceeding with hardly any regard for mechanistic concepts of the underlying defect structure.

Part of the difficulty stems from a peculiar lack of communication between the technologist and the basic researcher. The applied metallurgist frequently is not aware of the implications of new structural concepts; the theoretician is out of touch with practical problems. Another difficulty is that the defect theory has been in flux, and such advice as might be available for practical application is conflicting and of little direct aid. Finally, the methods of controlling and manipulating the new elements of structure through alloying and thermal and mechanical treatment must be devised before advanced concepts of structure can be effectively exploited.

REFERENCES

- (1) Bechtold, J. H. , Wessel, E. T. , and France, L. L. , Refractory Metals and Alloys, Interscience, New York (1961), p 25.
- (2) Bechtold, J. H. , Trans. AIME, 196, 1496 (1953).
- (3) Weaver, C. W. , J. Inst. Met. , 89, 385 (1960-61).
- (4) Bechtold, J. H. , and Shewmon, P. G. , Trans. A. S. M. , 46, 397 (1954).
- (5) Conrad, H. , and Hayes, W. , "Thermally Activated Deformation of the B. C. C. Metals at Low Temperatures" (Submitted to A. S. M. Trans.).
- (6) Conrad, H. , and Hayes, W. , "Correlation of the Thermal Component of the Yield Stress of the B. C. C. Metals" (Submitted to A. S. M. Trans.)
- (7) Schadler, H. W. , and Low, J. R. , "Low-Temperature Brittleness of Refractory Metals", O.N.R. Report on Contract Nonr-2614(00), (April, 1962).
- (8) Chaudhuri, A. R. , Patel, J. R. , and Rubin, L. G. , J. App. Phys. 33, 2736 (1962).
- (9) Reid, C. N. , Gilbert, A. , and Hahn, G. T. , (unpublished work).
- (10) Stein, D. , and Low, J. R. , Jr. , J. App. Phys. , 31 362 (1960).
- (11) Hahn, G. T. , Acta Met. , 10, 727 (1962).
- (12) Johnston, W. G. , J. App. Phys. , 33, 2716 (1962).
- (13) Hahn, G. T. , Gilbert, A. , and Reid, C. N. , (unpublished work).
- (14) Hahn, G. T. , Gilbert, A. , and Reid, C. N. , (submitted to J. Inst. Met.).

- (15) Scott, H. , Taebel, W. A. , and Lawthers, D. D. , The Metal Molybdenum, A. S.M. , Cleveland, Ohio, (1958), p 51.
- (16) Clark, J. W. , and Jones, E. S. , ASD TR 61-584 (January, 1962).
- (17) Lawley, A, van den Sype, J. , and Maddin, R. , J. Inst. Met. , 91, 23 (1962-63).
- (18) Fink, C. G. , Trans. Am. Electrochem Soc, 17, 229 (1910); and 22, 499 (1912).
- (19) Geach, G. A. , and Hughes, J. E. , 1955 Plansee Proceedings, p 245.
- (20) Allen, B. C. , Reid, C. N. , and Jaffee, R. I. , Battelle Report to NASA (March, 1962).
- (21) Klopp, W. D. , Holden, F. C. , and Jaffee, R. I. , Battelle Technical Report Nonr-1512(00) (July, 1960).
- (22) Hahn, G. T. , Gilbert, A. , and Jaffee, R. I. , 1962 Chicago Refractory Metals Symposium Interscience, New York (to be published).
- (23) Pugh, J. W. , Amra, L. H. , and Hurd, D. T. , Trans ASM, 55, 451 (1962).
- (24) Allen, B. C. , and Jaffee, R. I. , (to be published).
- (25) Bryant, R. T. , J. Less-Common Metals, 4, 62 (1962).
- (26) Jaffee, R. I. , Sims, C. T. , and Harwood, J. J. , 1958 Plansee Proceedings, p 380.
- (27) Allen, B. C. , Maykuth, D. J. , and Jaffee, R. I. , NASA TN D-837 (April, 1961).
- (28) Gilbert, A. , Reid, C. N. , and Hahn, (unpublished work).
- (29) Johansen, H. A. , Gilbert, H. L. , Nehar, R. G. , and Carpenter, R. L. , U. S. Bur. Mines Rept. Invest. No. 5056 (May, 1954).
- (30) Marcinkowski, M. J. , and Lipsitt, H. A. , Acta Met, 10, 95 (1962).
- (31) Pugh, S. W. , Trans ASM, 50, 1072 (1958).
- (32) Carreker, R. P. , and Guard, R. W. , Trans AIME, 206, 178 (1956).
- (33) Pugh, J. W. , Trans ASM, 47, 984 (1955).
- (34) Schnitzel, R. H. , WADD TR 60-37 Part III, (June, 1962).
- (35) Cottrell, A. H. , Dislocations and Plastic Flow in Crystals, Oxford University Press, (1953), p 136.
- (36) Schoeck, A. , and Seeger, A. , Acta Met, 7, 469 (1959).

- (37) Chang, W. H. , ASD TDR-62-211 (April, 1962).
- (38) Semchyshen, M. , The Metal Molybdenum, A. S. M. Cleveland, Ohio (1958), p 281.
- (39) Chang, W. H. , Perlmutter, I. , 1960 High Temperature Materials Conference, Cleveland, Ohio.
- (40) Jaffee, R. I. , The Metal Molybdenum, A. S. M. , Cleveland, Ohio (1958), p 330.
- (41) Allen, B. C. , 1961 Plansee Proceedings, p 770.
- (42) Atkinson, R. H. , Keith, G. H. , and Koo, R. C. , Refractory Metals and Alloys, Interscience, New York (1961), p 344.
- (43) Bunshaw, R. F. , and Goetzel, C. G. , WADC TR 59-414 (March, 1960).
- (44) Milner, Acta. Technica. Acad. Sci. Hang. , XVII, 67 (1957).
- (45) Mannerkoski, M. , Acta Met. , 10, 982 (1962).
- (46) Reick, G. D. , Phillips Tech. Rev. , 19, 113 (1957-1958).

RIJ/GTH:js

LIST OF DMIC TECHNICAL REPORTS ISSUED
DEFENSE METALS INFORMATION CENTER

Battelle Memorial Institute

Columbus 1, Ohio

Copies of the technical reports listed below may be obtained from DMIC at no cost by Government agencies, and by Government contractors, subcontractors, and their suppliers. Others may obtain copies from the Office of Technical Services, Department of Commerce, Washington 25, D. C. See PB numbers and prices in parentheses.

DMIC Report Number	Title
46D	Department of Defense Titanium Sheet-Rolling Program - Uniform Testing Procedure for Sheet Materials, September 12, 1958 (PB 121648 \$1.25)
46E	Department of Defense Titanium Sheet-Rolling Program - Thermal Stability of the Titanium Sheet-Rolling-Program Alloys, November 25, 1958 (PB 151061 \$1.25)
46F	Department of Defense Titanium Sheet-Rolling Program Status Report No. 4, March 20, 1959 (PB 151065 \$2.25)
46G	Department of Defense Titanium Sheet-Rolling Program - Time-Temperature-Transformation Diagrams of the Titanium Sheet-Rolling Program Alloys, October 19, 1959 (PB 151075 \$2.25)
46H	Department of Defense Titanium Sheet-Rolling Program, Status Report No. 5, June 1, 1960 (PB 151087 \$2.00)
46I	Statistical Analysis of Tensile Properties of Heat-Treated Ti-4A1-3Mo-1V Sheet, September 16, 1960 (PB 151095 \$1.25)
46J	Statistical Analysis of Tensile Properties of Heat-Treated Ti-4A1-3Mo-1V and Ti-2.5A1-16V Sheet (AD 259284 \$1.25)
106	Beryllium for Structural Applications, August 15, 1958 (PB 121648 \$3.00)
107	Tensile Properties of Titanium Alloys at Low Temperature, January 15, 1959 (PB 151062 \$1.25)
108	Welding and Brazing of Molybdenum, March 1, 1959 (PB 151063 \$1.25)
109	Coatings for Protecting Molybdenum From Oxidation at Elevated Temperature, March 6, 1959 (PB 151064 \$1.25)
110	The All-Beta Titanium Alloy (Ti-13V-11Cr-3Al), April 17, 1959 (PB 151066 \$3.00)
111	The Physical Metallurgy of Precipitation-Hardenable Stainless Steels, April 20, 1959 (PB 151067 \$2.00)
112	Physical and Mechanical Properties of Nine Commercial Precipitation-Hardenable Stainless Steels, May 1, 1959 (PB 151068 \$3.25)
113	Properties of Certain Cold-Rolled Austenitic Stainless Sheet Steels, May 15, 1959 (PB 151069 \$1.75)
114	Ductile-Brittle Transition in the Refractory Metals, June 25, 1959 (PB 151070 \$2.00)
115	The Fabrication of Tungsten, August 14, 1959 (PB 151071 \$1.75)
116R	Design Information on 5Cr-Mo-V Alloy Steels (H-11 and 5Cr-Mo-V Aircraft Steel) for Aircraft and Missiles (Revised), September 30, 1960 (PB 151072-R \$1.50)
117	Titanium Alloys for High-Temperature Use Strengthened by Fibers or Dispersed Particles, August 31, 1959 (PB 151073 \$2.00)
118	Welding of High-Strength Steels for Aircraft and Missile Applications, October 12, 1959 (PB 151074 \$2.25)
119	Heat Treatment of High-Strength Steels for Aircraft Applications, November 27, 1959 (PB 151076 \$2.50)
120	A Review of Certain Ferrous Castings Applications in Aircraft and Missiles, December 18, 1959 (PB 151077 \$1.50)
121	Methods for Conducting Short-Time Tensile, Creep, and Creep-Rupture Tests Under Conditions of Rapid Heating, December 20, 1959 (PB 151078 \$1.25)
122	The Welding of Titanium and Titanium Alloys, December 31, 1959 (PB 151079 \$1.75)
123	Oxidation Behavior and Protective Coatings for Columbium and Columbium-Base Alloys, January 15, 1960 (PB 151080 \$2.25)
124	Current Tests for Evaluating Fracture Toughness of Sheet Metals at High Strength Levels, January 28, 1960 (PB 151081 \$2.00)
125	Physical and Mechanical Properties of Columbium and Columbium-Base Alloys, February 22, 1960 (PB 151082 \$1.75)
126	Structural Damage in Thermally Cycled René 41 and Astroloy Sheet Materials, February 29, 1960 (PB 151083 \$0.75)
127	Physical and Mechanical Properties of Tungsten and Tungsten-Base Alloys, March 15, 1960 (PB 151084 \$1.75)
128	A Summary of Comparative Properties of Air-Melted and Vacuum-Melted Steels and Superalloys, March 28, 1960 (PB 151085 \$2.75)
129	Physical Properties of Some Nickel-Base Alloys, May 20, 1960 (PB 151086 \$2.75)
130	Selected Short-Time Tensile and Creep Data Obtained Under Conditions of Rapid Heating, June 17, 1960 (PB 151088 \$2.25)
131	New Developments of the Welding of Metals, June 24, 1960 (PB 151089 \$1.25)
132	Design Information on Nickel-Base Alloys for Aircraft and Missiles, July 20, 1960 (PB 151090 \$3.00)
133	Tantalum and Tantalum Alloys, July 26, 1960 (PB 151091 \$5.00)
134	Strain Aging of Refractory Metals, August 12, 1960 (PB 151092 \$1.75)
135	Design Information on PH 15-7 Mo Stainless Steel for Aircraft and Missiles, August 22, 1960 (PB 151093 \$1.25)

136A	The Effects of Alloying Elements in Titanium, Volume A. Constitution, September 15, 1960 (PB 151094 \$3.50)
136B	The Effects of Alloying Elements in Titanium, Volume B. Physical and Chemical Properties, Deformation and Transformation Characteristics, May 29, 1961 (AD 260256 \$3.00)
137	Design Information on 17-7 PH Stainless Steels for Aircraft and Missiles, September 23, 1960 (PB 151096 \$1.00)
138	Availability and Mechanical Properties of High-Strength Steel Extrusions, October 26, 1960 (PB 151097 \$1.75)
139	Melting and Casting of the Refractory Metals Molybdenum, Columbium, Tantalum, and Tungsten, November 18, 1960 (PB 151098 \$1.00)
140	Physical and Mechanical Properties of Commercial Molybdenum-Base Alloys, November 30, 1960 (PB 151099 \$3.00)
141	Titanium-Alloy Forgings, December 19, 1960 (PB 151100 \$2.25)
142	Environmental Factors Influencing Metals Applications in Space Vehicles, December 27, 1960 (PB 151101 \$1.25)
143	High-Strength-Steel Forgings, January 5, 1961 (PB 151102 \$1.75)
144	Stress-Corrosion Cracking - A Nontechnical Introduction to the Problem, January 6, 1961 (PB 151103 \$0.75)
145	Design Information on Titanium Alloys for Aircraft and Missiles, January 10, 1961 (PB 151104 \$2.25)
146	Manual for Beryllium Prospectors, January 18, 1961 (PB 151105 \$1.00)
147	The Factors Influencing the Fracture Characteristics of High-Strength Steel, February 6, 1961 (PB 151106 \$1.25)
148	Review of Current Data on the Tensile Properties of Metals at Very Low Temperatures, February 14, 1961 (PB 151107 \$2.00)
149	Bracing for High Temperature Service, February 21, 1961 (PB 151108 \$1.00)
150	A Review of Bending Methods for Stainless Steel Tubing, March 2, 1961 (PB 151109 \$1.50)
151	Environmental and Metallurgical Factors of Stress-Corrosion Cracking in High-Strength Steels, April 14, 1961 (PB 151110 \$0.75)
152	Binary and Ternary Phase Diagrams of Columbium, Molybdenum, Tantalum, and Tungsten, April 28, 1961 (AD 257739 \$3.50)
153	Physical Metallurgy of Nickel-Base Superalloys, May 5, 1961 (AD 258041 \$1.25)
154	Evolution of Ultrahigh-Strength, Hardenable Steels for Solid-Propellant Rocket-Motor Cases, May 25, 1961 (AD 257976 \$1.25)
155	Oxidation of Tungsten, July 17, 1961 (AD 263598 \$3.00)
156	Design Information on AM-350 Stainless Steel for Aircraft and Missiles, July 28, 1961 (AD 262407 \$1.50)
157	A Summary of the Theory of Fracture in Metals, August 7, 1961 (PB 151081 \$1.75)
158	Stress-Corrosion Cracking of High-Strength Stainless Steels in Atmospheric Environments, September 16, 1961 (AD 266005 \$1.25)
159	Gas-Pressure Bonding, September 25, 1961 (AD 265133 \$1.25)
160	Introduction to Metals for Elevated-Temperature Use, October 27, 1961 (AD 268647 \$2.50)
161	Status Report No. 1 on Department of Defense Refractory Metals Sheet-Rolling Program, November 2, 1961 (AD 267077 \$1.00)
162	Coatings for the Protection of Refractory Metals From Oxidation, November 24, 1961 (AD 271384 \$3.50)
163	Control of Dimensions in High-Strength Heat-Treated Steel Parts, November 29, 1961 (AD 270045 \$1.00)
164	Semiaustenitic Precipitation-Hardenable Stainless Steels, December 6, 1961 (AD 274805 \$2.75)
165	Methods of Evaluating Welded Joints, December 28, 1961 (AD 278088 \$2.25)
166	The Effect of Nuclear Radiation on Structural Metals, September 15, 1961 (AD 265839 \$2.50)
167	Summary of the Fifth Meeting of the Refractory Composites Working Group, March 12, 1962 (AD 274804 \$2.00)
168	Beryllium for Structural Applications, 1958-1960, May 18, 1962 (AD 276723 \$3.50)
169	The Effect of Molten Alkali Metals on Containment Metals and Alloys at High Temperatures, May 18, 1962 (AD 282932 \$1.50)
170	Chemical Vapor Deposition, June 4, 1962 (AD 281887 \$2.25)
171	The Physical Metallurgy of Cobalt-Base Superalloys, July 6, 1962 (AD 283356 \$2.25)
172	Background for the Development of Materials To Be Used in High-Strength-Steel Structural Weldments, July 31, 1962 (AD 284865 \$3.00)
173	New Developments in Welded Fabrication of Large Solid-Fuel Rocket-Motor Cases, August 6, 1962 (AD 284829 \$1.00)
174	Electron-Beam Processes, September 15, 1962 (AD 287433 \$1.75)
175	Summary of the Sixth Meeting of the Refractory Composites Working Group, September 24, 1962 (AD 287029 \$1.75)
176	Status Report No. 2 on Department of Defense Refractory Metals Sheet-Rolling Program, October 15, 1962 (AD 288127 \$1.25)
177	Thermal Radiative Properties of Selected Materials, November 15, 1962, Vol. I (AD 294345 \$3.00)
177	Thermal Radiative Properties of Selected Materials, November 15, 1962, Vol. II (AD 294346 \$4.00)
178	Steels for Large Solid-Propellant Rocket-Motor Cases, November 20, 1962
179	A Guide to the Literature on High-Velocity Metalworking, December 3, 1962
180	Design Considerations in Selecting Materials for Large Solid-Propellant Rocket-Motor Cases, December 10, 1962
181	Joining of Nickel-Base Alloys, December 20, 1962

<p>Battelle Memorial Institute, Defense Metals Information Center, Columbus, Ohio. STRUCTURAL CONSIDERATIONS IN DEVELOPING REFRACTORY METAL ALLOYS, by R. I. Jaffee and G. T. Hahn. January 31, 1963, 29 pp incl. illus., tables, refs. (DMIC Report 182) [AF 33(616)-7747] Unclassified report</p> <p>This report examines progress in applying advanced structural concepts to the problem of strengthening the Group VI-A metals, chromium, molybdenum, and tungsten. The role of dislocations and grain boundaries, dislocation-interstitial atom interactions, and stable dispersions at low, intermediate, and high temperatures is identified. These concepts provide a useful</p> <p>rationale of mechanical strength, but have not contributed significantly to practical developments.</p>	<p>Battelle Memorial Institute, Defense Metals Information Center, Columbus, Ohio. STRUCTURAL CONSIDERATIONS IN DEVELOPING REFRACTORY METAL ALLOYS, by R. I. Jaffee and G. T. Hahn. January 31, 1963, 29 pp incl. illus., tables, refs. (DMIC Report 182) [AF 33(616)-7747] Unclassified report</p> <p>This report examines progress in applying advanced structural concepts to the problem of strengthening the Group VI-A metals, chromium, molybdenum, and tungsten. The role of dislocations and grain boundaries, dislocation-interstitial atom interactions, and stable dispersions at low, intermediate, and high temperatures is identified. These concepts provide a useful</p> <p>rationale of mechanical strength, but have not contributed significantly to practical developments.</p>	<p>UNCLASSIFIED</p> <p>1. Heat-resistant alloys - development</p> <p>2. Heat-resistant alloys - metallurgy</p> <p>I. R. I. Jaffee</p> <p>II. G. T. Hahn</p> <p>III. Defense Metals Information Center</p> <p>IV. Contract AF 33(616)-7747</p> <p>UNCLASSIFIED</p>	<p>UNCLASSIFIED</p> <p>1. Heat-resistant alloys - development</p> <p>2. Heat-resistant alloys - metallurgy</p> <p>I. R. I. Jaffee</p> <p>II. G. T. Hahn</p> <p>III. Defense Metals Information Center</p> <p>IV. Contract AF 33(616)-7747</p> <p>UNCLASSIFIED</p>	<p>UNCLASSIFIED</p> <p>1. Heat-resistant alloys - development</p> <p>2. Heat-resistant alloys - metallurgy</p> <p>I. R. I. Jaffee</p> <p>II. G. T. Hahn</p> <p>III. Defense Metals Information Center</p> <p>IV. Contract AF 33(616)-7747</p> <p>UNCLASSIFIED</p>
--	--	---	---	---